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THESIS

Type 1 Hot Corrosion Furnace Testing
and Evaluation

by

Thomas Leon McGowen

October 1982

Thesis Advisor: Donald H. Boone

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data available from a variety of burner rig tests and will serve as a base line for further type 1 hot corrosion materials and coating research.

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Type 1 Hot Corrosion Furnace Testing and Evaluation

by

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Lieutenant, United States Navy
B.S., University of Kansas, 1975

Submitted in partial fulfillment of the
requirements for the degree of

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ABSTRACT

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I. INTRODUCTION AND BACKGROUND

The development of the gas turbine engine has been chiefly as a prime mover for aircraft in large part because of its inherently high thrust to weight ratio. Although the thermodynamic principles of gas turbine operation have been understood for decades, practical applications did not occur until the 1940's. This was a result of the development of alloys which could withstand the high temperature, high stress environment of the turbine hot section. These iron, nickel or cobalt alloys have since been classified as superalloys. Modern superalloys for gas turbine use are nickel or cobalt based. Principle superalloy development philosophy has been based on the criteria of sustained high power operation in a clean environment at altitude. These criteria are those applicable to aircraft operation.

The gas turbine offers many advantages as a prime mover in marine applications as well, especially for combatant ships. The fast startup and quick response, and the ability to efficiently burn many different fuel types are all desirable properties of gas turbines. The marine gas turbine also has the advantage of small volume and weight to horsepower ratios in addition to the reduced maintenance and down time associated with modular construction.

The United States Navy has committed itself to the use of the marine gas turbine for many of its new ships. The particular engine chosen for development and use by the Navy was the CF6/TF39 aircraft engine core used on the C5A transport aircraft. The marinized version of this engine has been designated the LM 2500. However, the LM 2500 engines in a marine environment have demonstrated significantly shorter life spans due to hot corrosion degradation than their counterparts in aircraft applications.

Low power operation and the aggressive marine environment of shipboard gas turbines provide new criteria for superalloy development. Although hot corrosion was a recognized phenomenon in aircraft operation, its increased severity in the marine environment was unexpected and required expanded research in alloy development. This presented a significant problem in that alloy compositions chosen to meet the mechanical criteria for gas turbine applications are often detrimental to corrosion resistance.

A. SUPERALLOYS

The superalloys are nickel or cobalt based with various other elements added to achieve high temperature creep and tensile strength, resistance to mechanical and thermal fatigue, as well as resistance to oxidation and hot corrosion. In nickel based alloys, aluminum and titanium are added to produce a gamma prime precipitate which provides the major

elevated-temperature strengthening mechanism. Additions of chromium provide grain boundary strengthening through formation of carbides and also provide effective solid solution strengthener [Ref. 1]. Aluminum and chromium both form protective oxides and their presence in superalloys can be beneficial for oxidation and hot corrosion resistance. Chromium also enters the gamma prime phase with the adverse effect of reducing the gamma prime solvus temperature. Therefore, alloys with high chromium contents will have a lower strength at high temperatures compared with alloys with lower chromium contents but containing other solid solution strengthening elements such as tungsten and molybdenum [Ref. 2]. Since alloys with high chromium contents have a high resistance to hot corrosion, attempts to increase high-temperature strength by decreasing the chromium content will normally reduce the hot corrosion resistance [Ref. 3].

In general, cobalt based superalloys are inherently more corrosion resistant than nickel based alloys. Their corrosion resistance is not completely understood but could in part be the result of the higher chromium content of cobalt alloys. Cobalt superalloys are lower in strength and have higher melting points than the nickel based alloys. X-40, a cobalt based superalloy, is used for the guide vanes in the LM 2500 engine where low stress and high temperatures are encountered.

B. COATINGS

Previous experience with gas turbine engines has shown that the use of protective coatings is mandatory even in the most benign operating environments [Ref. 4]. The purpose of a coating is to provide for the selective formation of a protective oxide suitable for the use environment. Generally, this oxide is Al_2O_3 but can be Cr_2O_3 as well.

In selecting the proper coating/substrate system for a particular application, the physical and mechanical properties of the coating must be similar or compatible to the substrate. For this reason the current coatings in use in gas turbine engines are either metallic overlay coatings or aluminide diffusion coatings. The uncoated alloys must also possess a degree of corrosion resistance because there are uncoated areas on the airfoil components and coating loss can occur during operation for several reasons. For example, some interior surfaces are difficult to coat and the coating may be removed by wear or FOD (foreign object damage).

Aluminum coatings are formed on superalloys by the pack cementation process which is a gaseous chemical vapor deposition and diffusion process. The resulting coating consists of an inner reaction-diffusion zone at the coating/substrate interface and one or two out zones consisting of intermetallic compounds of the metal aluminide type [Ref. 5]. Oxidation of the aluminide surface forms an adherent alumina

(Al_2O_3) scale which serves as an environmental barrier. If the alumina barrier is damaged by erosion or spalling during thermal cycling the underlying aluminide oxidizes to reform the alumina. This process can continue until the aluminum content of the coating is reduced to below the critical level for alumina formation.

By varying the activity of the aluminum in the pack cementation process and the deposition temperature, two general types of coatings can be applied. The coating on a nickel alloy produced in a high-activity pack consists mainly of an intermetallic phase based on Ni_2Al_3. With subsequent time at temperature in an inert environment aluminum will diffuse into the substrate from the Ni_2AL_3 phase resulting in the formation of an NiAl based phase. The coating produced in this way is referred to as an "inward" aluminide because of the initial inward diffusion of aluminum. Substrate elements and phases are present to approximately 70% in the outer layer of this three zone coating. The coating produced in the low-activity pack consists of the NiAl based phase only. The coating produced involves the outward diffusion of nickel from the substrate and is referred to as an "outward" aluminide. No substrate containing outer layer is present in this two zone coating. Typical inward and outward aluminide coating structures are shown in Figures 1 and 2.

Aluminide coatings are brittle at low and intermediate temperatures and provide only moderate hot corrosion resistance when compared to most overlay coatings. Continued development of aluminide coatings has resulted in duplex coatings or modified aluminides with enhanced corrosion resistance. This process involves the addition of elements such as chromium or noble metals such as platinum to the aluminide coating [Ref. 6]. The incorporation of these potentially beneficial elements can be accomplished by a number of techniques such as the inclusion of particles of the element in the pack or by predeposition prior to aluminizing. These developments plus the low cost and ease of application of the aluminides assure their continued use in the gas turbine industry.

The limitations of the aluminide coatings; brittleness, only moderate corrosion resistance, and strong substrate dependence, have led to the development of a new type of coating system, namely the metallic overlay. These coatings are known as "MCrAlY" alloy coatings (M=Fe,Ni, and/or Co) and are presently applied mainly by physical vapor deposition (PVD) processes [Ref. 7]. The alloys consist of two phases, a brittle aluminide phase in a ductile, chromium rich solid solution matrix. A typical overlay coating is shown in Figure 3. The coatings contain from 4% to 13% aluminum, 18% to 40% chromium, and 0.1% to 0.5% yttrium with the balance either cobalt and/or nickel. The chromium and aluminum are

protective oxide formers and the yttrium provides oxide adherence. The ability to vary the composition of these coatings for specific applications and their mechanical compatibility with the superalloy substrates are significant advantages over the aluminide coatings.

For example, the cobalt base BC-21 coating used on the first and second stage blades of the LM 2500 engine has a high chromium content, 20-24%, to increase hot corrosion resistance. This high chromium necessitates a lower aluminum content, 10-12%, to maintain sufficient ductility and adequate thermal fatigue. In contrast, the cobalt base ATD-6 (Airco Temescal) coating is used on the first stage blades of the AV8A Harrier engine. This blade requires a very ductile coating to combat the thermal fatigue caused by the high temperature cyclic operation of the engine. Accordingly, its composition is 18% chromium and 8% aluminum. As a result, the hot corrosion and oxidation resistance is less than that of BC 21 but the environment is less corrosive and the required life time is much less than that of the LM 2500.

Ceramic coatings are also being studied for use in gas turbines because they offer the dual advantage of superior corrosion resistance and the provision of an insulating thermal barrier. This barrier would allow increased turbine inlet temperature and/or reduced cooling air requirements. Presently, ceramic coatings have not been developed with

sufficiently compatible mechanical properties for application on critical superalloy airfoil substrates [Ref. 8].

Aluminide coatings are limited in their corrosion resistance due to the detrimental effects of incorporation of substrate elements in the coating. This has been well documented elsewhere [Ref. 9]. Initially, the overlay and ceramic coatings were thought to be substrate independent. However, it has been found that this is not so and there is a need to understand the interactions which occur in these systems. A research program at the Naval Postgraduate School has been concerned with the effect of substrate on the corrosion resistance of coatings. The program has focused on the type 2, low temperature corrosion resistance.

C. HOT CORROSION

Hot corrosion is a particularly aggressive, accelerated type of oxidation which attacks superalloys exposed to many types of gaseous combustion products. Hot corrosion of gas turbine blades and vanes occurs as a result of deposits of sulfates, primarily Na_2SO_4 originating from the intake air, and the presence of SO_2/SO_3, originating from the fuel. It has only recently been determined that there are two distinct types of hot corrosion in the gas turbine. Type 1 hot corrosion occurs in a temperature range above about 850°C. Type 2 hot corrosion occurs at lower temperatures, around 700°C. Corrosion between these temperatures is often a mixture of both types [Ref. 10].

Turbine inlet temperature is indicative of power level and since aircraft operation of gas turbines is almost exclusively at high power levels, type 1 hot corrosion has been recognized since the 1950's. Until recently, alloy and coating development efforts have been on the basis of aircraft gas turbines. In 1975, type 2 hot corrosion was recognized aboard the GTS CALLAGHAN, a test ship used to evaluate naval gas turbines and materials. This was a direct result of extended low power operation of the gas turbines aboard the CALLAGHAN. Consequently, substrate coating systems must be designed which have good corrosion resistance over the entire temperature range [Ref. 11].

The type 1 hot corrosion degradation morphology is shown schematically in Figure 4. Due to the molten salt film and the partial pressure gradients of O_2 and SO_3, the low partial pressure of O_2 at the molten salt/protective oxide interface causes the following reactions to occur:

$$2SO_4^{--} \rightarrow S_2 + 3O_2 + 2O^{--} \qquad 1$$

$$Al_2O_3 + O^{--} \rightarrow 2AlO_2^{-} \qquad 2(a)$$

and/or

$$2Cr_2O_3 + 3O_2 + 4O \rightarrow 4CrO_4^{--} \qquad 2(b)$$

This results in the dissolution of the protective oxides. As the AlO_2^{--} and/or CrO_4^{--} leave the interface and diffuse toward the surface of the melt, the partial pressures of O_2

and SO_3 increase and these ions reprecipitate as Al_2O_3 and Cr_2O_3. For an alloy containing aluminum which forms an Al_2O_3 protective oxide, reaction 2(a) occurs more readily than 2(b). Thus, aluminum is oxidized to Al_2O_3 and fluxed by reaction 2(a) resulting in a depletion zone under the oxide surface where the aluminum content is significantly reduced [Ref. 12]. For a MCrAlY coating this is seen by an absence of beta (MAl) phase to some depth below the oxide layer. For an uncoated alloy, the depletion layer is marked by a reduced amount of gamma prime just below the oxide surface. This aluminum depletion zone advances ahead of the corrosion front and is an identifying characteristic of type 1 hot corrosion.

Type 1 hot corrosion occurs above the melting temperature of NA_2SO_4 (886°C) and hot corrosion would not be expected to occur below that temperature. In type 2 hot corrosion it was found that an eutectic mixture of salts was produced on the blade surfaces. These eutectics, Na_2SO_4+MSO_4 (M=Co,Ni), have melting points as low as 575°C. The metal oxides are present in the protective oxide layer and the formation of the molten eutectic salt is the initiation phase of type 2 hot corrosion. This liquid salt can penetrate the oxide layer at cracks and results in attack of the alloy. Aluminum and sulfite ions react at the low partial pressures of SO_3 and O_2 to form $Al_2(SO_3)_3$ which later precipitates as Al_2O_3 in areas of higher partial

pressures of SO_3 and O_2. This process is called acid fluxing and results in the severe pitting associated with type 2 hot corrosion [Ref. 13]. Since this type of corrosion attacks the alloy at imperfections in the protective oxide layer, it can be much more severe than type 1 hot corrosion and harder to protect against. For less protective oxides, e.g. uncoated alloys, the attack is more general as in type 1 but with no aluminum depletion layer.

The sources of undesirable compounds necessary for hot corrosion, Na_2SO_4 and SO_3, are the intake air and fuel. Fuels available today are as free of contaminants as can be expected and afforded and may deteriorate in this respect in the years to come. Installation of demisters in the intakes of marine gas turbines has resulted in reduced hot corrosion through the reduction of salt spray ingestion. But it is unlikely that the ingestion can be eliminated in this way.

D. HOT CORROSION TESTING

In continuing the research effort at the Naval Postgraduate School, it was the purpose of this thesis to expand the hot corrosion testing capabilities by developing a furnace test which would reproduce type 1 hot corrosion. With this goal met, substrate effects on hot corrosion resistance of coatings can be studied in both type 1 and type 2 forms of hot corrosion.

There are many factors which affect the hot corrosion mechanisms and their rate of attack. The alloy composition, of course, is the primary variable which is studied in efforts to control attack. As the concentration of some elements varies, their effect can vary from beneficial to deleterious. As mentioned previously, the interactions between elements in the substrate/coating system are of particular interest. The fabrication condition of the alloy is also an important factor. Inhomogeneities, as a result of casting segregation for example, may produce localized hot corrosion attack. This attack will then spread to locations which are more corrosion-resistant [Ref. 14].

Gas composition encompasses two variables, partial pressures of O_2 and SO_3, which are critical to both the type of attack and the rate of attack. As can be seen from the stability diagram of Figure 4, the partial pressure gradients across the molten salt layer determine the degradation mechanism. Although type 1 and type 2 hot corrosion are identified by temperature range, it is possible to have type 2 hot corrosion at higher temperatures if sufficient SO_3 partial pressure is present. Efforts to accelerate type 1 hot corrosion by increasing SO_3 partial pressure is limited by this fact. The SO_3 partial pressure is also inversely related to temperature such that for a given SO_2 input rate, the SO_3 partial pressure can be increased by decreasing the temperature.

Gas velocity is also a corrosion rate parameter in the operating environment. Gas velocity controls the deposition rate of salts and may also cause erosion of the coating due to particles in the gas stream. Gas velocity is not considered significant (in laboratory tests performed) in the ranking of alloys for corrosion resistance.

Salt composition and deposition rate are the primary factors in the hot corrosion rate. A molten salt layer is necessary for hot corrosion. In the type 2 regime this requires the formation of an eutectic mixture. The thickness of the molten salt layer affects the partial pressure gradients. The salt deposition rate determines the length of time to initiation of hot corrosion and since salt is consumed in the corrosion process, the deposition rate must be sufficient to sustain the reaction.

Hot corrosion processes are dependent upon temperature. In the first case, the temperature must be sufficient to produce a molten salt layer. At higher temperature the liquid salt is less viscous and, therefore, the salt deposition rate decreases. As mentioned earlier, temperature is inversely related to SO_3 partial pressure. Thermal cycling can also result in cracking and spalling of the oxide and increased corrosion rates [Ref. 15].

Historically, hot corrosion testing and evaluations have been accomplished using burner rigs with gas velocities up to mach 1.0. These rigs were developed to simulate

conditions in operating gas turbines. By adding sulfur to the fuel and an aqueous spray of salts to the intake air, and ducting the combustion products to eliminate their dilution before interaction with the sample, hot corrosion can be achieved. This produces an environmentally realistic test. Burner rigs are of two types; the high pressure rig most nearly simulates the engine environment but is expensive and does not allow precise control of the individual parameters, and the similar low pressure burner rig is less expensive but does not allow any better control and is less relevant to the engine environment. It should be noted that results from the two types of rigs are not significantly different.

The necessity for understanding the basic mechanisms of hot corrosion led to development of laboratory furnace tests. These tests incorporate only the significant variables necessary to produce the phenomenon: temperature, SO_3 partial pressure, salt amount, and composition. A furnace test which not only meets this requirement but also provides a simple means of assessing the hot corrosion resistance of materials, as does the burner rig, has been developed and is the major concern of this thesis. Furnace tests are significantly less expensive, often of shorter duration and allow precise control of the specific parameters. These factors make the furnace test more suited to university programs. The tube furnace design used at the Naval Postgraduate School is shown in Figure 5.

Actual engine testing is, of course, the ultimate test particularly in terms of lifetime predictions. But rapid screening tests such as the burner rig and furnace tests are necessary for the evaluation of the number of variables in the environment and alloy/substrate systems and to test predictions made from theories.

II. EXPERIMENTAL PROCEDURE

A laboratory tube furnace at the Naval Postgraduate School has been in use since October 1979 and has successfully reproduced type 2 hot corrosion on both coated and uncoated alloys [Ref. 16]. Since the primary variables necessary to produce hot corrosion are the same for both type 1 and 2 temperature regimes this furnace was used without modification for testing in the type 1 hot corrosion temperature range.

The furnace is a resistance type tube furnace with a 2-3/8 inch inside diameter ceramic tube insert, and end caps. Temperature in the furnace can be maintained with $\pm$ 10°C in the hot zone of the furnace, approximately six inches, where the test specimens are placed. A thermocouple is the sensor for the proportional controller and a digital pyrometer is used to monitor temperature.

Compressed air is regulated and passed through moisture indicating "drierite" desicant at a rate of 2000 ml/min. The air is mixed with SO_2 at a flow rate of 20 ml/min to produce a mixture containing 1% by volume and gas velocity of 1 cm/sec over the test specimens. The air mixture enters the furnace through a stainless steel tube and flows the length of the furnace in the tube to preheat the gas mixture prior to flowing back over the specimens. The air is finally exhausted into a fume hood.

The primary variables in the tube furnace test are temperature, percent SO_2 gas, amount of $NaSO_4$ coating, cycle time and total exposure time. Using IN 100 as a standard, because of its poor corrosion resistance, several short term test runs were made varying these parameters to determine the values which would produce the most aggressive type 1 attack. From these results given in Figure 6, 900°C and 1% SO_2 were chosen as test values.

Since the sulfate coating is applied prior to thermal exposure, salt deposition rate is eliminated as a parameter. For this investigation, the amount of Na_2SO_4 coating was not considered a significant variable as long as sufficient salt was present. Under this assumption, a cycle time of 50 hours was chosen with approximately 2 mg/cm^2 of Na_2SO_4 applied at the beginning of each cycle. This regimen is more convenient than a 20 hour cycle time with 1 mg/cm^2 of Na_2SO_4 in the extended testing under type 1 conditions.

Total exposure time was limited to 500 hours, ten 50 hour cycles. This limit was based on data produced in various burner rig tests of 500 hour duration. By eliminating the parameter of salt deposition rate from the tube furnace test, it was thought that more severe corrosion would be produced.

Test specimens were chosen from alloy and alloy/coating systems which could be used to compare results with data available from various burner rig tests. Test specimens are listed in Table 2 and the alloy compositions are listed in

Table 3. Most specimens were pin type specimens approximately 3 cm long and 0.63 cm in diameter. Rectangular coupons were also used where pins were not available.

The specimens were visually inspected, weighed and their dimensions recorded. They were then preheated to 150°C, sprayed with a saturated Na_2SO_4 solution and weighed again. This procedure was repeated until a nominal weight gain equivalent to 2 mg/cm^2 of Na_2SO_4 was accumulated. The specimens were then placed in the tube furnace hot zone on a specimen holder made of Al_2O_3 base fire brick. After 50 hours, the specimens were removed from the furnace, air cooled, washed, visually examined, and resalted. This procedure was continued until a total exposure time of 500 hours was accumulated. The exception to this procedure was the IN 100 sample which was removed from the test after 100 hours and microscopically examined to determine if type 1 hot corrosion was occurring.

Following furnace exposure, the pins were sectioned and prepared for microscopic examination using standard metallographic procedures. Corrosion morphology and severity of attack were determined. Depth of attack was measured as recommended by Aprigliano [Ref. 17]. Selected specimens were examined under a scanning electron microscope and backscatter images were taken for cobalt, chromium, and aluminum to further verify type 1 degradation morphology.

III. RESULTS AND DISCUSSION

A. FURNACE PARAMETERS

In developing a type 1 hot corrosion furnace test, one of the goals was to accelerate the corrosion rate compared with burner rigs, while maintaining type 1 hot corrosion morphology. The selection of furnace parameters listed in Table 3 was based upon the plot of Corrosion Factors vs. Corrosion Rate shown in Figure 5. This graph was constructed from a minimum of data and the inherent scatter associated with this type of testing precludes anything but the most general conclusions at this time. Also, extrapolation of the results for a single alloy with poor corrosion resistance to other alloy and alloy/coating systems may not always be valid.

Figure 6 indicated an increase in corrosion rate with increased volume percent SO_2, as predicted. To assure type 1 hot corrosion morphology in the tests, 1% SO_2 was the maximum value used even though the corrosion rate should be more aggressive above this value.

A maximum in corrosion rate vs. temperature is also seen in Figure 6. This maximum occurring at approximately 900°C is important in lending credence to the tests. The peak is associated with the most viscous, adherent form of the molten Na_2SO_4 layer. At temperatures below 900°C, pure Na_2SO_4 is solid (T_m=886°C) and at temperatures above 900°C the salt becomes less viscous and does not adhere to the surface [Ref. 18].

Figure 6 provides no direct correlation between corrosion rate and cycle time. It would be expected that increased corrosion would occur with shorter more numerous cycles. The IN 100 used for these tests produced a very thick, heavy scale and the oxides formed were not protective. With more corrosion resistant alloys, and particularly coatings, the effects of cycle time will vary with the tenacity of the oxide formed. A 50 hour cycle time was chosen for use in this study for further testing to maintain simplicity in the test.

The test parameters listed in Table 3 for the furnace test were selected from the above observations. These parameters are also consistent with the parameters used in burner rig evaluations [Refs. 19, 20, 21], which will be used to correlate the relative severity of the furnace test.

B. REPRODUCTION OF TYPE 1 HOT CORROSION

Upon completion of the 500 hour furnace test of the alloys and alloy/coating systems listed in Table 1, the specimens were examined to verify type 1 hot corrosion as the mode of the attack and ranked by their relative corrosion resistance. Microscopic examination of a standard CoCrAlY specimen (IN 738 with BC 21 coating) revealed the occurrence of type 1 hot corrosion. A single phase aluminum-depleted layer beneath the oxide scale forms within the coating with type 1 hot corrosion of CoCrAlY. This single phase layer

appears lighter due to the loss of aluminum to the oxide scale from the darker aluminum-rich beta (CoAl). Sulfides will also form in this depletion zone. Figure 7 shows the presence of an aluminum depleted single phase band in the CoCrAlY (BC 21) coated IN 738 specimen in this test. Sulfides are also present in this layer.

To further verify the aluminum depletion zone indicative of type 1 hot corrosion, electron back-scatter images of the CoCrAlY specimen were taken under a scanning electron microscope. Figures 8 through 10 are examples of the corrosion morphology observed in other specimens tested. All specimens underwent what appeared to be type 1 hot corrosion degradation.

C. ALLOY SELECTION AND RANKING

Common alloy and alloy/coating systems were chosen for this research so that the results obtained could be compared with similar burner rig type 1 hot corrosion tests. A summary of this data produced by burner rig testing at David W. Taylor Naval Ship Research and Development Center [Ref. 22], Pratt & Whitney Aircraft Group [Ref. 23], and Detroit Diesel Allison [Ref. 24] is given in Table 4. The test parameters used in these various tests are listed in Table 3. In addition to the uncoated and CoCrAlY coated alloys, a series of aluminide coated samples were included in the test to provide baseline information for future research at the Naval Postgraduate School related to aircraft gas turbine operation in marine environments.

There is a significant variation in the data reported in Table 4. Variations between factors of 2 and 20 can be seen in the same alloy tested by the same facility. This run-to-run variation is inherent in burner rig testing. The corrosion parameters in the furnace test can be more finely controlled which should lead to reduced scatter in the results. However, the validity of this prediction awaits further testing.

In comparing the corrosion rates in the furnace test and the burner rig tests, the CoCrAlY coated alloys were the most corrosion resistant in all the tests. The CoCrAlY coatings were developed specifically for type 1 hot corrosion resistance [Ref. 25] and should be the most resistant. It was also found that the furnace test produced a more severe corrosion rate in the CoCrAlY coated alloys than any of the burner rig tests. However, this increase was not enough to classify the furnace test as an accelerated test.

Corrosion of the uncoated IN 738 was in close agreement with the amount of corrosion observed in IN 738 in the DDA burner rig test [Ref. 26]. The uncoated Rene 80 exhibited a wide variation in the amount of corrosion produced in the various burner rigs. In the Naval Postgraduate School furnace test the Rene 80 underwent a very aggressive selective form of corrosion at the 90° cut edge and was removed from the test after only 250 hours of exposure. Following the first 50 hour cycle heavy corrosion was observed at the top of the

specimen. As the test continued the corrosion moved down the specimen until less than 50% of the base metal was left after 250 hours. The selective progression of this corrosion could be the result of a combination of the substrate molybdenum content which promotes a basic fluxing mode of corrosion and specimen geometry but further investigation is necessary to understand this phenomenon. Since this corrosion was not indicative of typical type 1 hot corrosion, depth of corrosion readings was not taken.

The results of the furnace test correlate with those obtained from the burner rigs in type and amount of corrosion produced for these alloys and alloy/coating systems. Thus, this procedure provides for a simplified, though not accelerated, test for further studies and alloy screening.

The corrosion of the IN 738 aluminides tested varied from poor to excellent. The simple aluminides failed in under 500 hours of exposure. The corrosion rates obtained for both the inward and outward aluminides were as severe as in the uncoated IN 738, but are not truly indicative of the hot corrosion resistance of these aluminide coatings. Since the coatings were penetrated and consumed relatively early, the corrosion of the inner diffusion zone and substrate cannot be directly correlated with corrosion of an uncoated alloy. The inward aluminide coating had an original thickness of 4 mils and was penetrated after 300 hours of exposure. The higher aluminum content outward aluminide was originally 2

mils thick and was not penetrated until after 350 hours of exposure. In view of this, the outward aluminide was more corrosion resistant than the inward aluminide coating under type 1 conditions.

A single phase platinum aluminide coating and a two phase platinum aluminide [Ref. 27] were also exposed in the test. The two phase Pt-aluminide failed at 500 hours in the test which indicates better corrosion resistance than the simple aluminides. The single phase Pt-aluminide was more corrosion resistant, comparing favorably with the corrosion resistance of CoCrAlY. Because of the complex corrosion behavior of these modified aluminides, it would be simplistic at this time to draw detailed conclusions from the corrosion rates obtained. There is little doubt, however, that the single phase Pt-aluminide performed well in the test.

IV. CONCLUSIONS AND RECOMMENDATIONS

Based on the results of the furnace testing and evaluations carried out, the following conclusions can be made:

1. Type 1 hot corrosion can be reproduced by the resistance furnace test developed at the Naval Postgraduate School.
2. The corrosion rates achieved by the furnace test are comparable with those obtained in burner rig tests for both coated and uncoated superalloys.
3. Under the constraints of the furnace parameters chosen, the corrosion rates can be varied but an accelerated test was not achieved.
4. Platinum can be effective in improving the type 1 hot corrosion resistance of aluminide coated superalloys. This platinum effect is highly dependent upon the coating application process and resulting structure.

The above conclusions suggest the following recommendations for further study:

1. Continue furnace testing with greater SO_2/SO_3 partial pressures to determine if significant acceleration of type 1 hot corrosion can be achieved while maintaining type 1 degradation morphology.
2. Conduct type 1 hot corrosion testing and evaluation of modified aluminides with control of the coating variables to further understand the role of these factors in their hot corrosion resistance.

3. Examine substrate effects observed for CoCrAlY coatings under type 2 conditions in this higher temperature, less agressive regime.

APPENDIX A

TABLE 1

Test Specimens

Alloy	Coating
IN 100	-
Rene 80	-
Rene 80	BC 21
IN 738	-
IN 738	BC 21
IN 738	inward aluminide
IN 738	outward aluminide
IN 738	one phase[1] aluminide
IN 738	two phase[1] aluminide

1 platinum modified

TABLE 2

Chemical Compositions of Alloys and CoCrAlY Coatings

Alloy	Alloy Elements (wt %)										
	Ni	C	Cr	Co	W	Mo	Al	Ti	B	Zr	Other
IN 100	59.5	0.18	10.0	15.0	--	3.0	5.5	4.7	0.014	0.06	1.0V
1N 738	61.4	0.17	16.0	8.5	2.6	1.75	3.4	3.4	0.010	0.10	1.75Ta, 0.9Nb
Rene 80	60.3	0.17	14.0	9.5	4.0	4.0	3.0	5.0	0.015	0.03	--
BC 21[1]	--	--	20.9	Bal	--	--	9.4	--	--	--	0.29Y
BC 21[2]	--	--	23.5	Bal	--	--	11.0	--	--	--	0.39Y

1 IN 738 specimen

2 Rene 80 specimen

TABLE 3

Test Parameters

Type of Test	Temp.	Salt	SO_2	Thermal Cycle	Duration
NPS Furnace Test	900°C	2 mg/cm^2 Na_2SO_4	1% in dry air	50 hrs	500 hrs
Burner Rig					
D W Taylor Naval Ship R&D Center	900°C	ASTM sea salt 10 ppm in air	marine diesel w/ 1% sulfur	24 hrs	500 hrs
Detroit Diesel Allison	899°C	ASTM sea salt 10 ppm in air	marine diesel w/ 1% sulfur	24 hrs	500 hrs
Pratt and Whitney	900°C	ASTM sea salt 20 ppm in air	1.3% added at combustor	20 hrs	500 hrs

TABLE 4

Corrosion Test Results

Alloy	Coating	Corrosion Rate (mils/1000 hrs)			
		Test Facility			
		NSRDC[1]	P&W[2]	DDA[3]	NPS
Rene 80	none	21.8	55.4	11.0	-
		50.0	-	3.6	-
		1.7	-	-	-
IN 738	none	-	-	12.8	11.8
		-	-	16.3	-
Rene 80	BC 21	1.8	-	1.2	2.4
		1.6	-	0.2	-
		0.7	-	-	-
IN 738	BC 21	-	-	-	3.2
	inward aluminide				21.0*
	outward aluminide				15.8*
	platinum aluminide[4]				23.6*
	platinum aluminide[5]				4.0

1 David W. Taylor Naval Ship Research and Development Center
2 Pratt and Whitney
3 Detroit Diesel Allison
4 two phase aluminide
5 single phase aluminide

* coating penetrated, includes some substrate

APPENDIX B

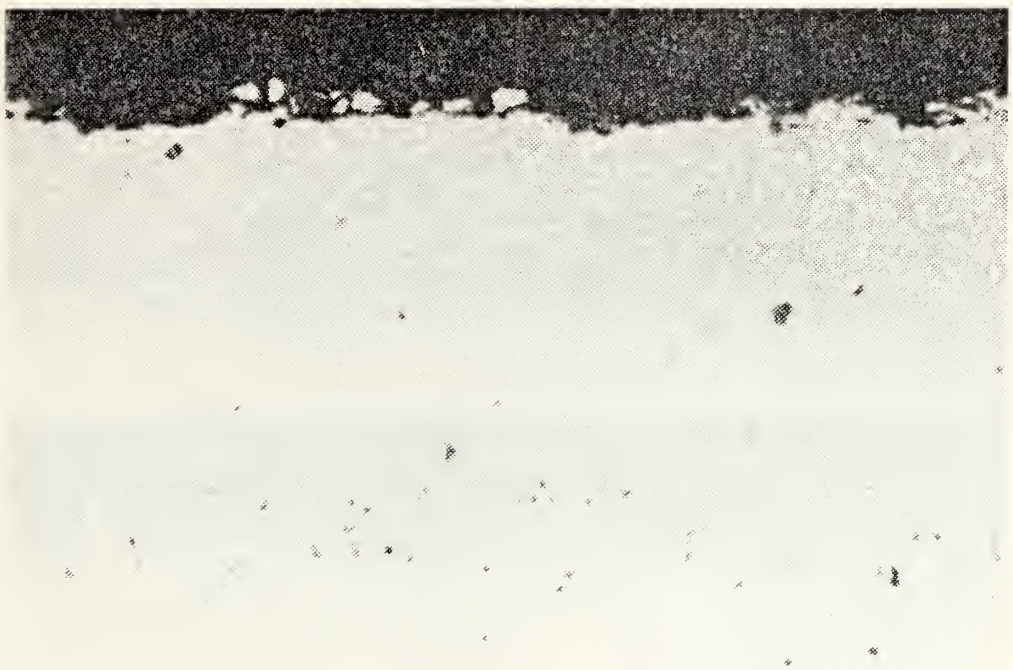


Figure 1. Typical Inward Aluminide Coating (650X).

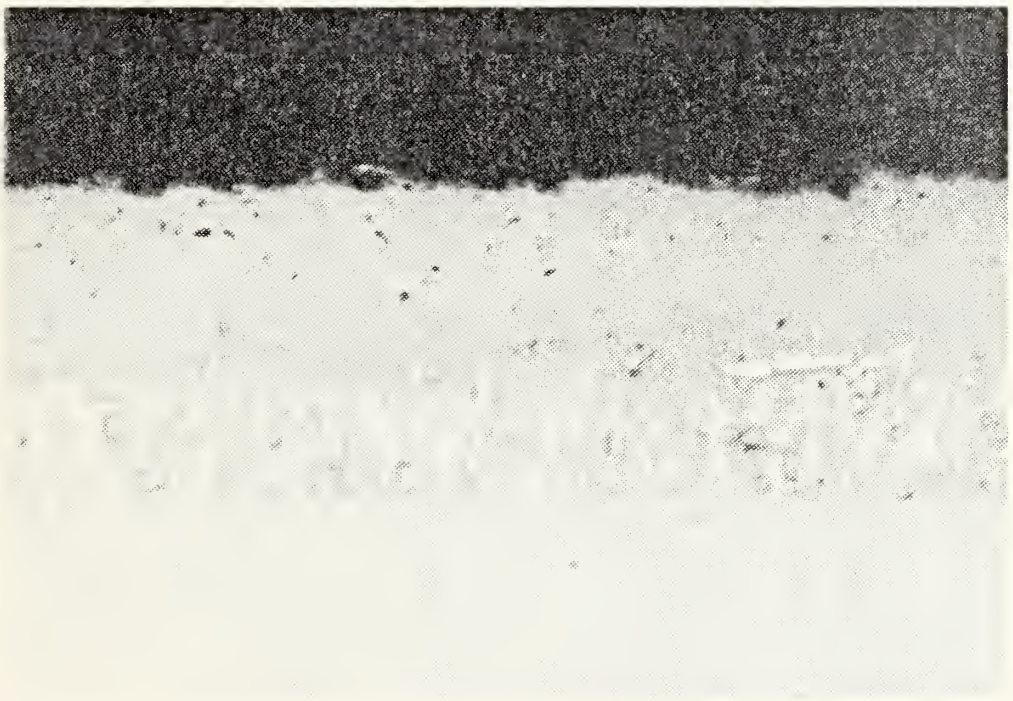


Figure 2. Typical Outward Aluminide Coating (650X).

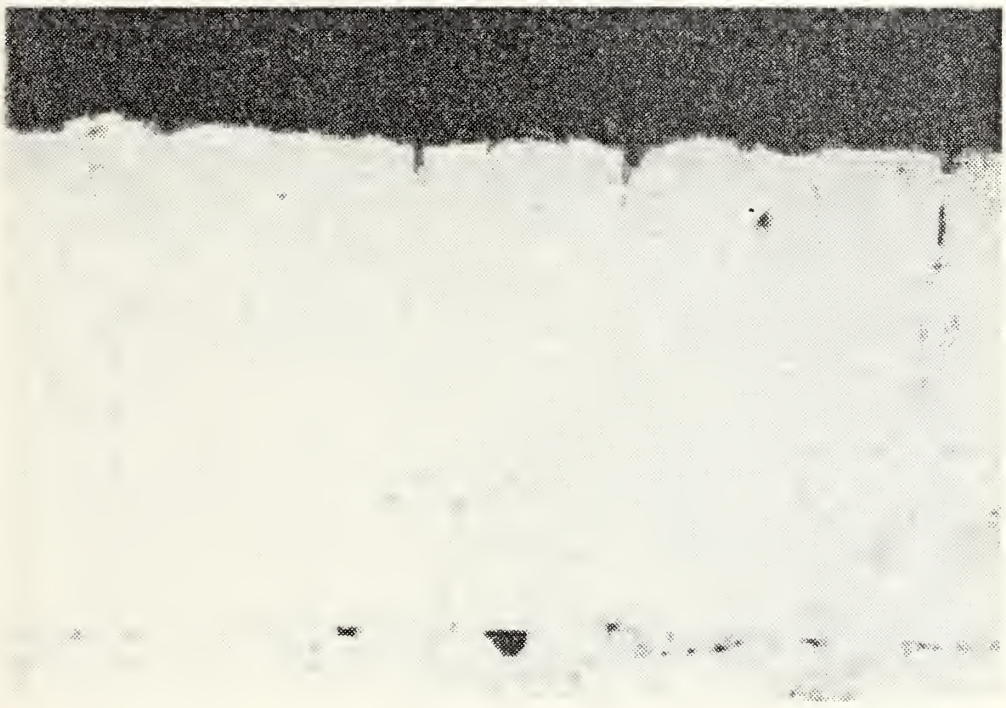
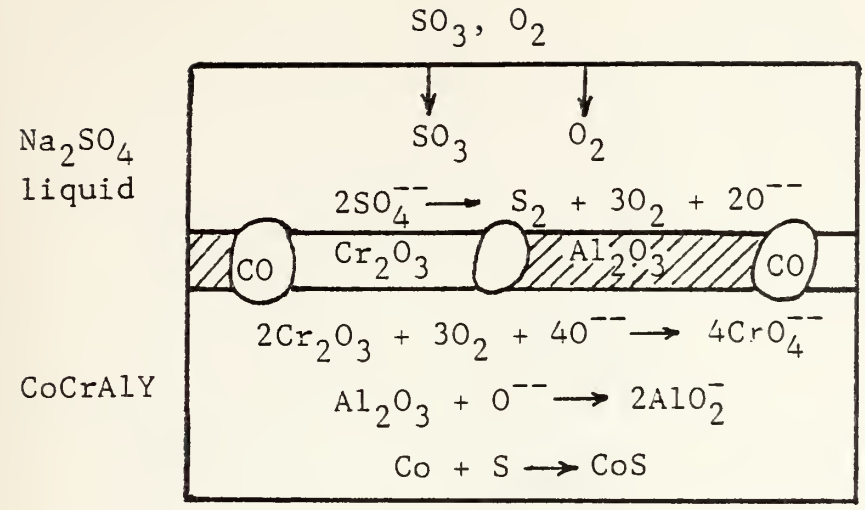
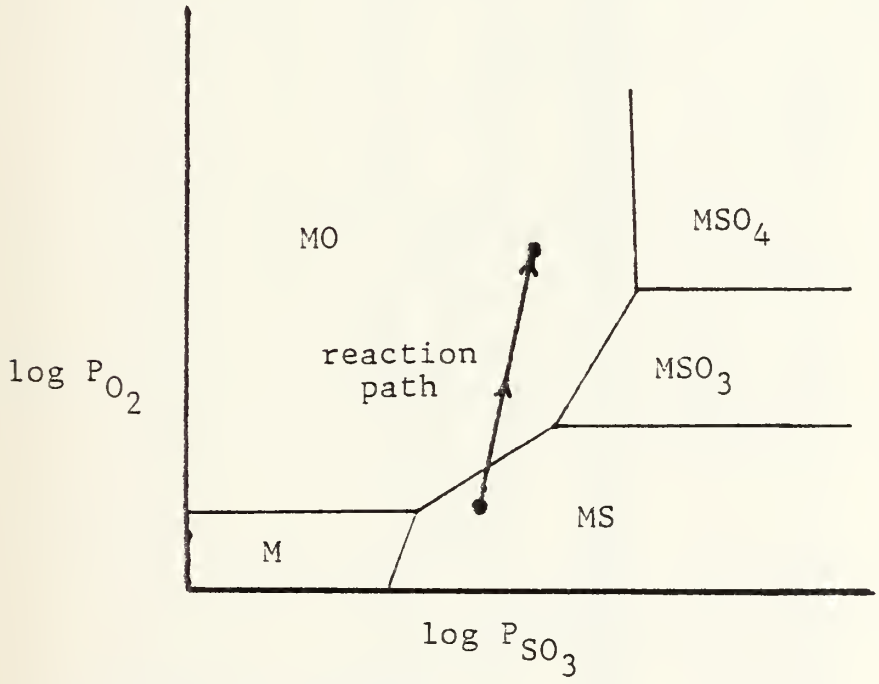


Figure 3. Typical CoCrAlY (BC 21) Overlay Coating, 650X, unetched.



(a) Type 1 hot corrosion reactions



(b) Stability Diagram

Figure 4. Schematic diagram to illustrate type 1 hot corrosion

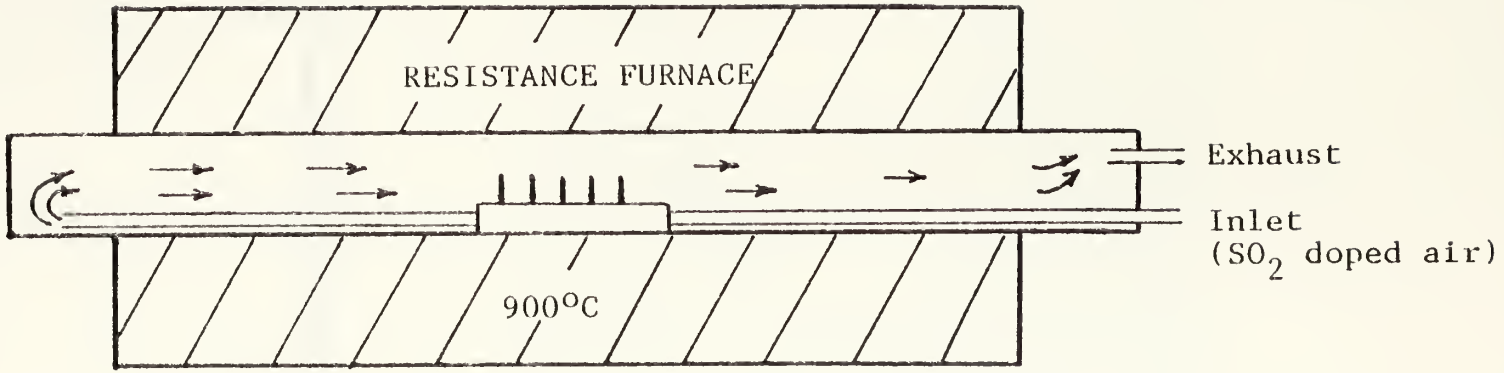


Figure 5. Cross Section of a Tube Furnace.

CORROSION RATES VS. CORROSION FACTORS

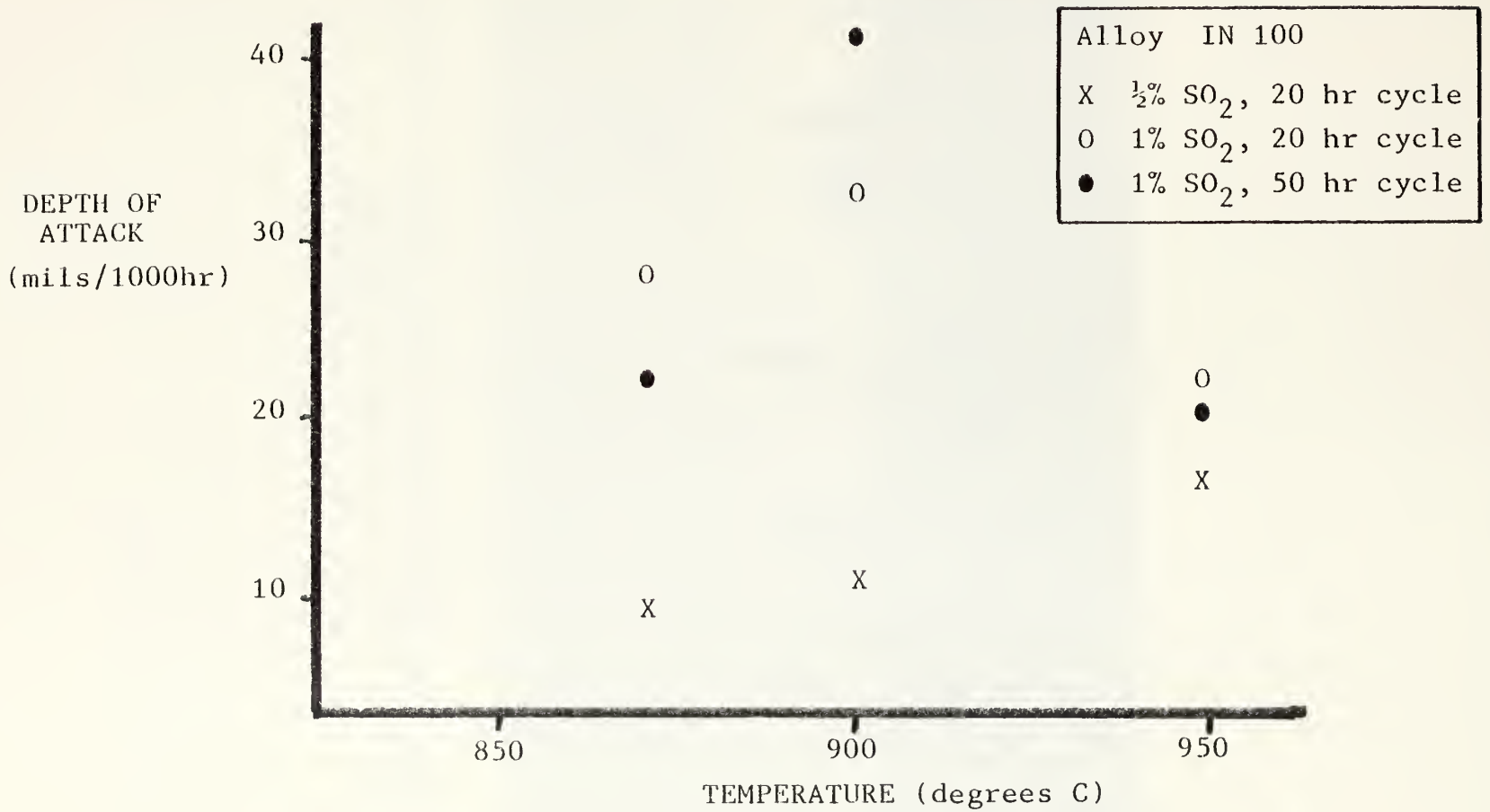


Figure 6. Corrosion Rates of Uncoated IN 100 with varying Furnace Test Parameters.

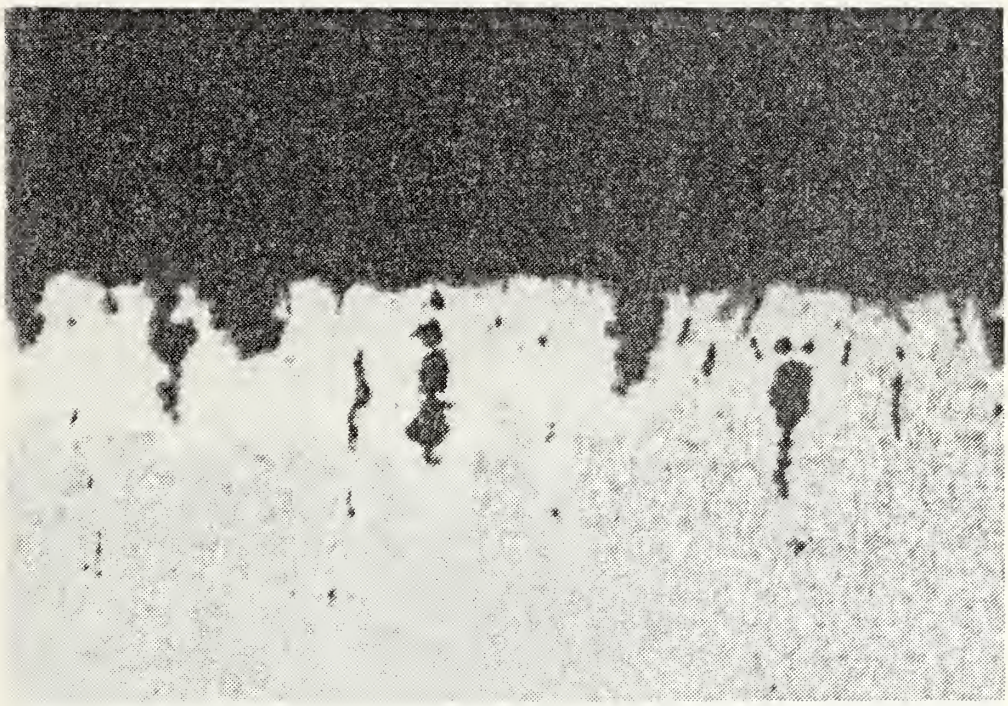


Figure 7. Type 1 Hot Corrosion in CoCrAlY (BC 21) Coating (500 Hours at 900° C), 650X, unetched.

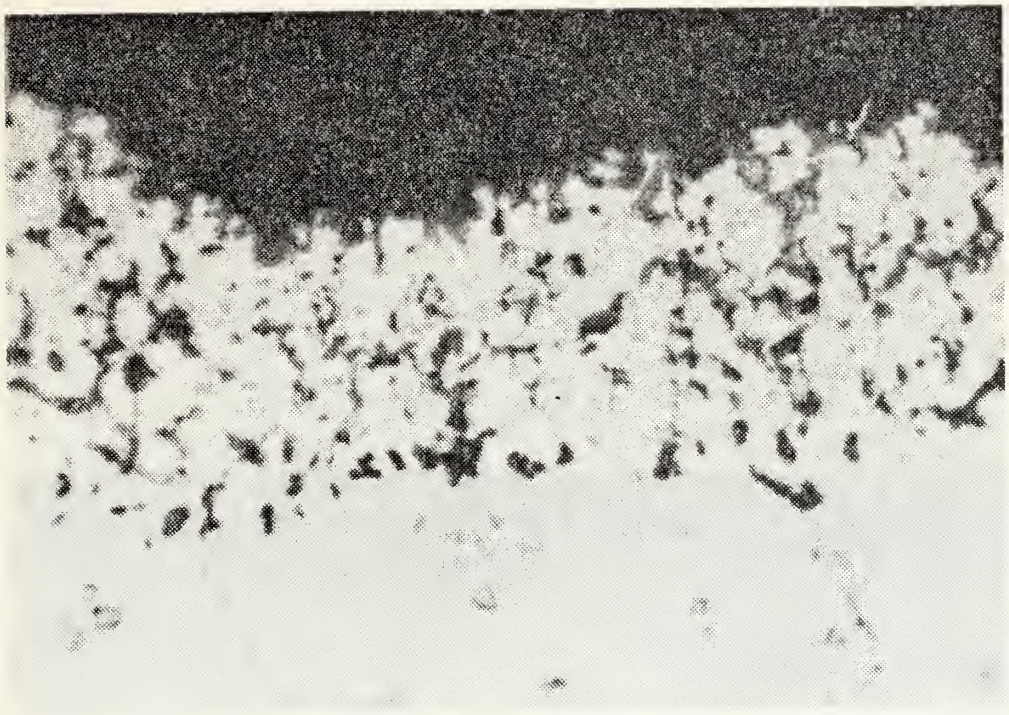


Figure 8. Type 1 Hot Corrosion in Uncoated IN 100 (500 Hours at 900°C), 650X, unetched.

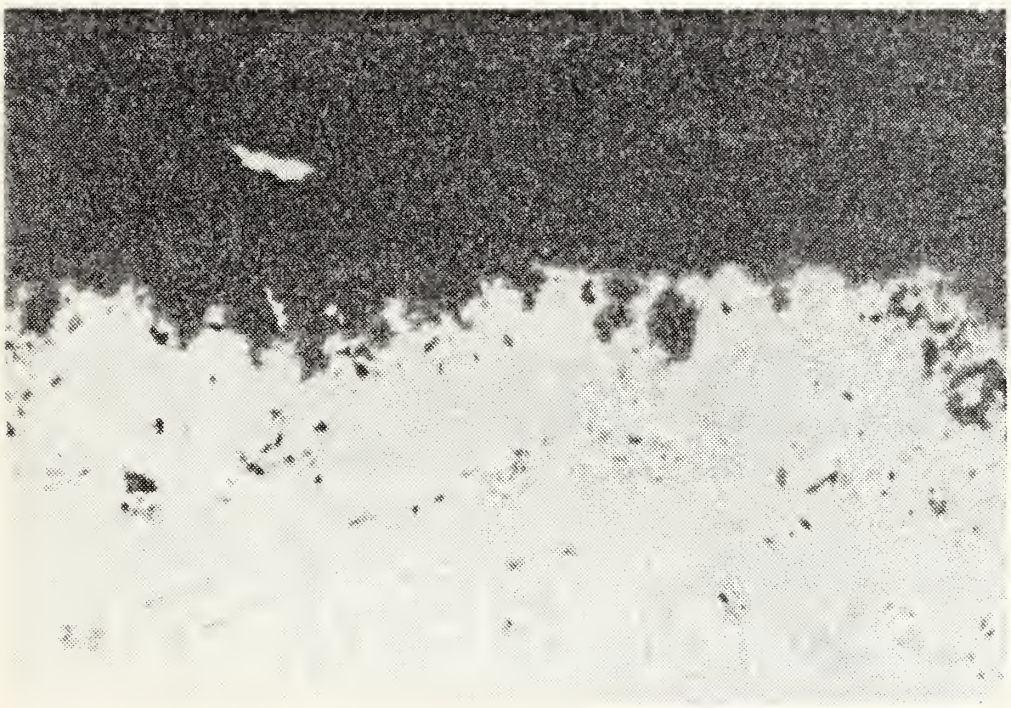


Figure 9. Type 1 Hot Corrosion in a Single Phase Platinum Aluminide Coating (500 Hours at 900°C), 650X, unetched.

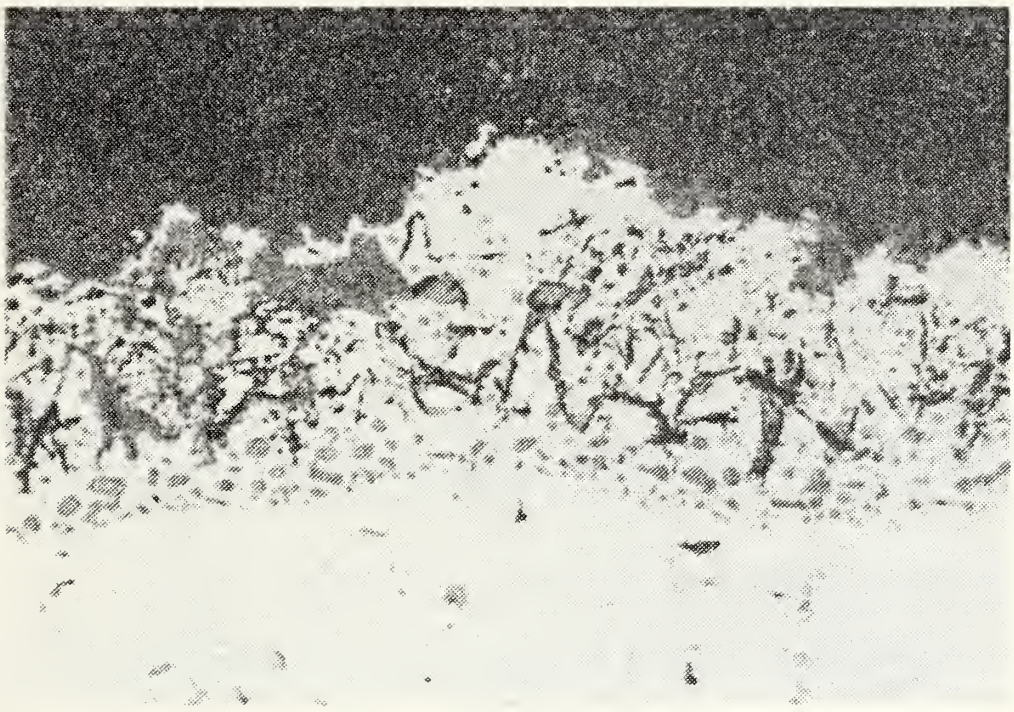


Figure 10. Type 1 Hot Corrosion in IN 738 Substrate after Outward Aluminide Coating was Consumed (500 Hours at 900°C), 650X, unetched

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